

**BAKE-HARDENABLE COLD ROLLED STEEL SHEET
HAVING EXCELLENT FORMABILITY,
AND METHOD OF MANUFACTURING THE SAME**

【Technical Field】

5 The present invention relates to cold rolled steel sheets for automobile bodies, and the like. More particularly, the present invention relates to bake-hardenable cold rolled steel sheets, improved in bake hardenability and formability by controlling a content of solid solution carbon in crystal grains with fine precipitates, and a method of manufacturing the same.

10 **【Background Art】**

 As for materials for exterior plates of automobile bodies, bake-hardening cold rolled steel sheets are generally used in order to improve dent resistance. The bake-hardening cold rolled steel sheets have excellent ductility through press forming, and increased yield strength through paint baking or coating treatment after press forming.
15 That is, as carbon or nitrogen is in solid solution in the steel as interstitial elements, and fixes dislocations created upon press forming, the yield point of the bake-hardening cold rolled steel sheets is increased.

 The bake-hardening cold rolled steel sheets include aluminum-killed steels, which are batch-annealed materials, and interstitial free steels (IF steels).

20 In the case of the aluminum-killed steels, which are batch-annealed materials, small amounts of solid solution carbon remain in the steel, and ensure aging resistance while providing bake hardenability in the order of 10 ~ 20 MPa after the baking treatment. However, for such batch-annealing materials, there are disadvantages of a lower increase in the yield strength even after the baking treatment, and of a lower
25 productivity.

 In the case of the IF steels, as solid solution carbon or nitrogen in the steel is completely precipitated by adding titanium or niobium to the steel, the formability of the steel is enhanced. Bake hardening IF steels are manufactured by imparting the bake

hardenability to the IF steels. For the bake hardened IF steels, the bake hardenability can be ensured by allowing an appropriate amount of carbon to remain in the steel through control of an added amount of titanium or niobium and an added amount of carbon. However, in the case of the bake hardening IF steels, in order to allow the
5 appropriate amount of carbon to remain in the solid solution in the steel, it is necessary to control the added amount of sulfur and nitrogen, which can react with titanium or niobium and create precipitates thereof, within a very narrow range, as well as the added amounts of carbon, titanium or niobium. Accordingly, it is difficult to ensure high quality products, and manufacturing costs are increased.

10 **【Disclosure】**
 【Technical Problem】

Therefore, the present invention has been made in view of the above problems, and it is an object of the present invention to provide bake-hardenable cold rolled steel sheets, which are improved in bake hardenability and formability due to a higher
15 plasticity-anisotropy index and a lower in-plane anisotropy index without adding Ti and Nb, and a method of manufacturing the same.

【Technical Solution】

In accordance with one aspect of the present invention, the above and other objects can be accomplished by the provision of a bake-hardenable cold rolled steel
20 sheet, comprising: 0.003 ~ 0.005 % of C; 0.003 ~ 0.03 % S; 0.01 ~ 0.1 % of Al; 0.02 % or less of N; 0.2 % or less of P; at least one of 0.03 ~ 0.2 % of Mn and 0.005 ~ 0.2 % of Cu; and the balance of Fe and other unavoidable impurities, in terms of weight%, wherein, when the steel sheet comprises one of Mn and Cu, a composition of Mn, Cu, and S satisfies one of the relationships: $0.58 * \text{Mn} / \text{S} \leq 10$ and $1 \leq 0.5 * \text{Cu} / \text{S} \leq 10$ in terms
25 of weight, and when the steel sheet comprises both Mn and Cu, a composition of Mn, Cu, and S satisfies the relationships: $\text{Mn} + \text{Cu} \leq 0.3$ and $2 \leq 0.5 * (\text{Mn} + \text{Cu}) / \text{S} \leq 20$ in terms of weight, and wherein precipitates of MnS, CuS, and (Mn, Cu)S have an average size of 0.2 μm or less.

The cold rolled steel sheet of the present invention can be classified into three types in accordance with added elements selected from the group consisting of Mn and Cu. That is, (1) Mn solely-added steel (Cu excluded, which will also be referred to as “MnS-precipitated steel”), (2) Cu solely-added steel (Mn excluded, which will also be referred to as “CuS-precipitated steel”), and (3) Mn and Cu added steel (which will also be referred to as “MnCu-precipitated steel”), which will be described in detail as follows.

(1) The MnS-precipitated steel comprises: 0.003 ~ 0.005 % of C; 0.005 ~ 0.03 % of S; 0.01 ~ 0.1 % of Al; 0.02 % or less of N; 0.2 % or less of P; 0.05 ~ 0.2 % of Mn; and the balance of Fe and other unavoidable impurities, in terms of weight%, wherein the composition of Mn and S satisfies the relationship: $0.58 \cdot \text{Mn}/\text{S} \leq 10$ in terms of weight, and wherein precipitates of MnS have an average size of 0.2 μm or less. The method of manufacturing MnS-precipitated steel comprises the steps of: hot-rolling a steel slab with finish rolling at an Ar_3 transformation temperature or more to provide a hot rolled steel sheet, after reheating the steel slab to a temperature of 1,100 $^{\circ}\text{C}$ or more, the steel slab comprising: 0.003 ~ 0.005 % of C; 0.005 ~ 0.03 % of S; 0.01 ~ 0.1 % of Al; 0.02 % or less of N; 0.2 % or less of P; 0.05 ~ 0.2 % of Mn; and the balance of Fe and other unavoidable impurities, in terms of weight%, wherein the composition of Mn and S satisfies the relationship: $0.58 \cdot \text{Mn}/\text{S} \leq 10$ in terms of weight; cooling the hot rolled steel sheet at a cooling rate of 200 $^{\circ}\text{C}/\text{min}$ or more; winding the cooled steel sheet at a temperature of 700 $^{\circ}\text{C}$ or less; cold rolling the steel sheet; and continuous annealing the cold rolled steel sheet.

(2) The CuS-precipitated steel comprises: 0.003 ~ 0.005 % of C; 0.003 ~ 0.025 % of S; 0.01 ~ 0.08 % of Al; 0.02 % or less of N; 0.2 % or less of P; 0.01 ~ 0.2 % of Cu; and the balance of Fe and other unavoidable impurities, in terms of weight%, wherein the composition of Cu and S satisfies the relationship: $1 \leq 0.5 \cdot \text{Cu}/\text{S} \leq 10$ in terms of weight, and wherein precipitates of CuS have an average size of 0.1 μm or less. The method of manufacturing CuS-precipitated steel comprises the steps of: hot-rolling a steel slab with finish rolling at an Ar_3 transformation temperature or more to provide a hot rolled steel sheet, after reheating the steel slab to a temperature of 1,100 $^{\circ}\text{C}$ or more, the steel slab comprising 0.003 ~ 0.005 % of C; 0.003 ~ 0.025 % of S; 0.01 ~ 0.08 % of Al; 0.02 % or less of N; 0.2 % or less of P; 0.01 ~ 0.2 % of Cu; and

the balance of Fe and other unavoidable impurities, in terms of weight%, wherein the composition of Cu and S satisfies the relationship: $1 \leq 0.5 * \text{Cu}/\text{S} \leq 10$ in terms of weight; cooling the hot rolled steel sheet at a cooling rate of 300 °C/min or more; winding the cooled steel sheet at a temperature of 700 °C or less; cold rolling the steel sheet; and continuous annealing the cold rolled steel sheet.

(3) The MnCu-precipitated steel comprises: 0.003 ~ 0.005 % of C; 0.003 ~ 0.025 % of S; 0.01 ~ 0.08 % of Al; 0.02 % or less of N; 0.2 % or less of P; 0.03 ~ 0.2 % of Mn; 0.005 ~ 0.2 % of Cu; and the balance of Fe and other unavoidable impurities, in terms of weight%, wherein the composition of Mn, Cu, and S satisfies the relationships: $\text{Mn} + \text{Cu} \leq 0.3$ and $2 \leq 0.5 * (\text{Mn} + \text{Cu})/\text{S} \leq 20$ in terms of weight, and wherein MnS, CuS, and (Mn, Cu)S precipitates have an average size of 0.2 μm or less. The method of manufacturing MnCu-precipitated steel comprises the steps of: hot-rolling a steel slab with finish rolling at an Ar₃ transformation temperature or more to provide a hot rolled steel sheet, after reheating the steel slab to a temperature of 1,100 °C or more, the steel slab comprising: 0.003 ~ 0.005 % of C; 0.003 ~ 0.025 % of S; 0.01 ~ 0.08 % of Al; 0.02 % or less of N; 0.2 % or less of P; 0.03 ~ 0.2 % of Mn; 0.005 ~ 0.2 % of Cu; and the balance of Fe and other unavoidable impurities, in terms of weight%, wherein the composition of Mn, Cu, and S satisfies the relationships: $\text{Mn} + \text{Cu} \leq 0.3$ and $2 \leq 0.5 * (\text{Mn} + \text{Cu})/\text{S} \leq 20$ in terms of weight; cooling the hot rolled steel sheet at a cooling rate of 300 °C/min; winding the cooled steel sheet at a temperature of 700 °C or less; cold rolling the steel sheet; and continuous annealing the cold rolled steel sheet.

The above bake-hardenable cold rolled steel sheet of the present invention may be applied to ductile cold rolled steel sheets having a 240 MPa-grade tensile strength or to high strength cold rolled steel sheets having a 340 MPa-grade or more tensile strength.

In the case of the ductile cold rolled steel sheets in a 240 MPa-grade, the steel sheet comprises 0.003 ~ 0.005 % of C, 0.003 ~ 0.03 % of S; 0.01 ~ 0.1 % of Al; 0.004 % or less of N; 0.015 % or less of P; at least one of 0.03 ~ 0.2 % of Mn and 0.005 ~ 0.2 % of Cu; and the balance of Fe and other unavoidable impurities, in terms of weight%, wherein, when the steel sheet comprises one of Mn and Cu, the composition of Mn, Cu, and S satisfies one of the relationships: $0.58 * \text{Mn}/\text{S} \leq 10$ and

$1 \leq 0.5 * \text{Cu/S} \leq 10$ in terms of weight, and when the steel sheet comprises both Mn and Cu, the composition of Mn, Cu, and S satisfies the relationship: $\text{Mn} + \text{Cu} \leq 0.3$ and $2 \leq 0.5 * (\text{Mn} + \text{Cu}) / \text{S} \leq 20$ in terms of weight, and wherein the precipitates of MnS, CuS, and (Mn, Cu)S have an average size of $0.2 \mu\text{m}$ or less.

5 In the case of the high strength cold rolled steel sheets in a 340 MPa-grade or more, it can be classified into steel, which contains at least one of P, Si, and Cr, as a solid solution-intensifier, and steel which contains a higher content of N, as a precipitation-intensifier. That is, it is preferred that at least one of 0.2 % or less of P, 0.1 ~ 0.8 % of Si, and 0.2 ~ 1.2 % of Cr be contained in the ductile cold rolled steel
10 sheet. If P is solely added to the ductile cold rolled steel sheet, the content of P is preferably in the range of 0.03 ~ 0.2 %. Alternatively, high strength characteristics can be ensured by means of AlN precipitates by increasing the content of N to 0.005 ~ 0.02 %, and adding 0.03 ~ 0.06 % of P.

15 In order to further enhance the formability of the cold rolled steel sheet, the steel sheet may further comprise 0.01 ~ 0.2 % of Mo.

【Advantageous Effects】

As apparent from the above description, according to the present invention, the bake-hardenable cold rolled steel sheets allows the content of solid solution in the crystal grains to be controlled by means of fine MnS, CuS, (Mn, Cu)S precipitates,
20 thereby providing improved bake hardenability, formability, yield strength, and yield strength-ductility balance.

【Description of Drawings】

The above and other objects, features and other advantages of the present invention will be more clearly understood from the following detailed description taken
25 in conjunction with the accompanying drawings, in which:

Figs. 1a to 1c are graphical representations showing the relationship between the content of solid solution carbon in crystal grains and the size of precipitates, in which Fig. 1a shows the case of MnS-precipitated steel, Fig. 1b shows the case of CuS-

precipitated steel, and Fig. 1c shows the case of MnCu-precipitated steel;

Figs. 2a and 2b are graphical representations showing the relationship between the size of MnS precipitate and the cooling rates, in which Fig. 2a shows the case of $0.58 \cdot \text{Mn}/\text{S} < 10$, and Fig. 2b shows the case of $0.58 \cdot \text{Mn}/\text{S} > 10$;

5 Figs. 3a to 3c are graphical representations showing the relationship between the size of CuS precipitates and the cooling rates, in which Fig. 3a shows the case of $0.5 \cdot \text{Cu}/\text{S} = 2.56$, Fig. 3b shows the case of $0.5 \cdot \text{Cu}/\text{S} = 8.1$, and Fig. 3c shows the case of $0.5 \cdot \text{Cu}/\text{S} = 28$; and

10 Figs. 4a and 4b are graphical representations showing the relationship between the size of MnS, CuS and (Mn, Cu)S precipitates and the cooling rates, in which Fig. 4a shows the case of $2 \leq 0.5 \cdot (\text{Mn} + \text{Cu})/\text{S} \leq 20$, and Fig. 4b shows the case of $0.5 \cdot (\text{Mn} + \text{Cu})/\text{S} > 20$.

【Best Mode】

15 Preferred embodiments of the present invention will now be described in detail. However, it should be noted that the present invention is not limited to these embodiments. For the composition of the steel sheets of the present invention, “weight %” will be simply represented as “%” hereinafter.

20 The inventors of the present invention have discovered new facts, as will be described below, through investigations into enhancement of bake hardenability without adding Ti and Nb. That is, the content of solid solution carbon in crystal grains can be appropriately controlled with fine precipitates of MnS, CuS, or (Mn, Cu)S, thereby increasing the yield strength, specifically the post-bake yield strength. These precipitates may have positive effects not only on an increase of the yield strength caused by precipitation strengthening, but also on the plasticity-anisotropy index and the in-plane anisotropy index.

25 Referring to Figs. 1a to 1c, it can be appreciated that the finer the distribution of MnS, CuS, and (Mn, Cu)S precipitates the greater the reduction in content of the solid solution carbon in the crystal grains. This is attributed to the relatively free movement of the solid solution carbon remaining in the crystal grain. More specifically, the solid solution carbon can move freely in the crystal grains, and be then

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easily coupled to movable dislocations, thereby influencing aging characteristics at room temperature. The solid solution carbon can also be moved to more stable regions, such as grain boundaries or around the precipitations, segregated in these regions, and activated therein at a high temperature, for example, during paint baking treatment, thereby influencing the bake hardenability. Accordingly, reduction in content of the solid solution carbon in the crystal grains means that carbon exists in the more stable region, such as the grain boundaries or around the fine precipitates, and influences the bake hardenability.

Figs. 1a to 1c are graphical representations showing the relationship between the content of solid solution carbon in crystal grains and the size of precipitates, in which Fig. 1a shows the case of MnS-precipitated steel, Fig. 1b shows the case of CuS-precipitated steel, and Fig. 1c shows the case of MnCu-precipitated steel. The finer the size of the precipitates the greater the reduction in the content of the solid solution carbon in the crystal grains, and for the carbon, which does not exist in the crystal grains among the total content of the carbon, it effectively influences the bake hardenability. Referring to in Figs. 1a to 1c, it can be seen that the content of the solid solution carbon in the crystal grains is reduced to approximately 20 ppm or less, when the MnS precipitates have a size of about 0.2 μm or less (Fig. 1a), when the CuS precipitates have a size of about 0.1 μm or less (Fig. 1b), and when the MnCu-precipitates have a size of about 0.2 μm or less (Fig. 1c).

As such, in order to secure the carbon content effective for the bake hardenability, it is important to have fine distribution of MnS, CuS or MnCu-precipitates while keeping the total content of carbon within the range of 0.003 ~ 0.005 % in the steel. Accordingly, there have been investigations into a method for finely distributing these precipitates, and the results indicate that the fine distribution of these precipitates in the crystal grains can be achieved by controlling the contents of Mn, Cu, and S, and the composition of these elements in the steel, together with controlling a cooling rate after hot rolling.

Fig. 2a is a graphical representation obtained after investigating the size of precipitates according to the cooling rate after hot rolling a steel sheet, which comprises: 0.004 % C; 0.15 % Mn; 0.008 % P; 0.015 % S; 0.03 % Al; and 0.0012 % N (where $0.58 \cdot \text{Mn}/\text{S} = 5.8$). Referring to Fig. 2a, the MnS precipitates have a size of

0.2 μm or less by controlling the cooling rate of the steel sheet under the condition wherein the combination of Mn and S satisfies the relationship: $0.58 \cdot \text{Mn}/\text{S} \leq 10$.

Fig. 3a is a graphical representation obtained after investigating the size of precipitates according to the cooling rate after hot rolling a steel sheet, which
5 comprises: 0.004 % C; 0.01 % P; 0.008 % S; 0.05 % Al; 0.0014 % N; and 0.041 % Cu (where $0.5 \cdot \text{Cu}/\text{S} = 2.56$). Referring to Fig. 3a, the CuS precipitates have a size of 0.1 μm or less by controlling the cooling rate of the steel sheet under the condition wherein the combination of Cu and S satisfies the relationship: $0.5 \cdot \text{Cu}/\text{S} \leq 10$.

Fig. 4a is a graphical representation obtained after investigating the size of
10 precipitates according to the cooling rate after cold rolling a steel sheet, which comprises: 0.004 % C; 0.13 % Mn; 0.009 % P; 0.015 % S; 0.04 % Al; 0.0029 % N; and 0.04 % Cu (where $\text{Mn} + \text{Cu} = 0.17$ % and $0.5 \cdot (\text{Mn} + \text{Cu})/\text{S} = 5.67$). Referring to Fig. 4a, the MnS, CuS, (Mn, Cu)S precipitates have a size of 0.2 μm or less by controlling the cooling rate of the steel sheet under the condition wherein the combination of Mn,
15 Cu, and S satisfies the relationship: $2 \leq 0.5 \cdot (\text{Mn} + \text{Cu})/\text{S} \leq 20$.

The bake-hardenable cold rolled steel sheet in accordance with the invention has high yield strength, and thus allows a reduction in thickness of the steel sheet. As a result, the cold rolled steel sheet in accordance with the invention has an effect of weight reduction for the products thereof. Moreover, the low in-plane anisotropy of
20 the cold rolled steel sheet of the invention minimizes formation of wrinkles or ears during or after processing of the steel sheet. The cold rolled steel sheet of the invention also has grain boundaries reinforced due to an appropriate content of the carbon remaining in the grain boundaries by the fine precipitates, thereby preventing the brittleness fracture caused by grain boundaries weakened after processing.

25 The bake-hardenable cold rolled steel sheet of the present invention, and a method of manufacturing the same will be described in detail as follows.

[Cold rolled steel sheet of the invention]

Carbon (C): 0.003 ~ 0.005 %

30 Although the content of solid solution carbon in crystal grains is reduced due to fine precipitates, since carbon is segregated in grain boundaries or surroundings of the fine precipitates, the bake hardenability is increased without damaging aging

characteristics at room temperature as a higher content of carbon is segregated in the grain boundaries or around the fine precipitates. That is, as the carbon content in the steel sheet increases, the increased carbon content is segregated in the grain boundaries or around the fine precipitates, which is an effective way to increase the bake
5 hardenability of the steel sheet. As a result, the carbon content must be 0.003 % or more in the steel sheet in order to ensure the bake hardenability. That is, in order to enhance the bake hardenability, the carbon content exceeds 0.0030 % and 0.0031 % or more. However, if the carbon content exceeds 0.005 %, the formability can be rapidly reduced. Accordingly, the carbon content is preferably in the range of 0.003 ~
10 0.005 %.

Sulfur (S): 0.003 ~ 0.03 %

A sulfur content less than 0.003 % can lead to not only decrease in amount of MnS, CuS and (Mn, Cu) precipitates, but also creation of excessively coarse
15 precipitates, thereby lowering the bake hardenability of the steel sheet. A content of sulfur more than 0.03 % can lead to a large amount of solid solution sulfur, thereby remarkably decreasing the ductility and the formability of the steel sheet, and increasing the possibility of hot shortness. According to the present invention, for the MnS-precipitated steel, the sulfur content is preferably in the range of 0.005 ~ 0.03 %, and for the CuS-precipitated steel, the sulfur content is preferably in the range of 0.003
20 ~ 0.025 %. For the MnCu-precipitated steel, the sulfur content is preferably in the range of 0.003 ~ 0.025 %.

Aluminum (Al): 0.01 ~ 0.1 %

Aluminum is an alloying element generally used as a deoxidizing agent.
25 However, in the present invention, aluminum is added to the steel for precipitating nitrogen in the steel, and thus prevents the formability from being lowered due to solid solution nitrogen. An aluminum content of less than 0.01 % can lead to an increase in content of solid solution nitrogen, thereby lowering the formability, whereas an
30 aluminum content of more than 0.1 % can lead to an increase of solid solution aluminum, thereby lowering the ductility of the steel. According to the present invention, for the CuS-precipitated steel and the MnCu-precipitated steel, the aluminum

content is preferably in the range of 0.01 ~ 0.08 %. If the nitrogen content is increased to 0.005 ~ 0.02%, a high strength steel sheet can be obtained by virtue of strengthening effects of AlN precipitates.

5 Nitrogen (N): 0.02 % or less

Nitrogen is an unavoidable element introduced into the steel during the steel manufacturing process, and in order to obtain strengthening effects, it is preferably added to the steel in an amount not exceeding 0.02 %. In order to obtain a ductile steel sheet, the nitrogen content is preferably 0.004 % or less. In order to obtain a
10 high strength steel sheet, the nitrogen content is preferably 0.005 ~ 0.02 %. Although the nitrogen content must be 0.005 % or more in order to obtain strengthening effects, a nitrogen content of more than 0.02 % leads to deterioration in formability of the steel sheet. In order to provide the high strength steel using nitrogen, a phosphorus content is preferably 0.03 ~ 0.06 %. According to the present invention, in order to ensure
15 high strength by virtue of the AlN precipitates, the combination of Al and N, that is, $0.52 \cdot \text{Al}/\text{N}$ (where Al and N are denoted in terms of wt%) is preferably in the range of 1 ~ 5. The combination of Al and N ($0.52 \cdot \text{Al}/\text{N}$) of less than 1 can lead to lowering the formability due to the solid solution nitrogen, whereas the combination of Al and N ($0.52 \cdot \text{Al}/\text{N}$) exceeding 5 leads to negligible strengthening effects.

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Phosphorus (P): 0.2 % or less

Phosphorus is an alloying element, which can increase solid solution strengthening effects while allowing a slight reduction in r-value (plasticity-anisotropy index), and can ensure the high strength of the steel in which the precipitates are
25 controlled. Accordingly, in order to ensure the high strength by controlling the content of P, the P content is preferably 0.2 % or less. A phosphorus content of more than 0.2 % can lead to a reduction in ductility of the steel sheet. When phosphorous alone is added to the steel in order to ensure the high strength of the steel sheet, the P content is preferably 0.03 ~ 0.2 %. For the ductile steel sheet, the P content is
30 preferably 0.015 % or less. For the steel sheet in which the high strength is ensured by use of the AlN precipitates, the P content is preferably 0.03 ~ 0.06 %. This is attributed to the fact that, although a phosphorus content of 0.03 % or more enables a

target strength to be ensured, a phosphorus content exceeding 0.06 % can lower the ductility and formability of the steel. According to the present invention, in the case where the high strength of the steel sheet is ensured by addition of Si and Cr, the P content can be appropriately controlled within 0.2 wt% or less in order to obtain the target strength. In this case, even if the P content is 0.015 % or less, the high strength can be ensured.

In the present invention, at least one of manganese (Mn) and copper (Cu) is preferably added to the steel. These elements are combined with sulfur (S), and create the MnS, CuS or (Mn, Cu)S precipitates.

Manganese (Mn): 0.03 ~ 0.2 %

Manganese is an alloying element, which precipitates the solid solution sulfur in the steel as the MnS precipitates, thereby preventing the hot shortness caused by the solid solution sulfur. In the present invention, Mn is precipitated as the fine MnS and/or (Mn, Cu)S precipitates under appropriate conditions for the combination of S and/or Cu with Mn and for the cooling rate. The fine precipitates can impart the bake hardenability to the steel sheet during the paint baking treatment by causing carbon to be segregated in the grain boundaries or around the precipitates rather than the crystal grains. In order to achieve these effects, the Mn content must be 0.03 % or more. Meanwhile, a manganese content exceeding 0.2 % causes coarse precipitates due to a higher content of manganese, thereby deteriorating the bake hardenability of the steel sheet. If Mn alone is added to the steel (excluding Cu), the manganese content is preferably 0.05 ~ 0.2 %.

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Copper (Cu): 0.005 ~ 0.2 %

Copper is an alloying element, which creates fine precipitates under appropriate conditions for the combination of S and/or Mn with Cu, and the cooling rate before a winding process in a hot rolling process. The fine precipitates can impart the bake hardenability to the steel sheet during the paint baking treatment by causing carbon to be segregated in the grain boundaries or around the precipitates rather than the crystal grains. In order to achieve these effects, the Cu content must be 0.005 %

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or more. Meanwhile, a copper content greater than 0.2 % causes coarse precipitates due to a higher content of copper, thereby deteriorating the bake hardenability of the steel sheet. If Cu alone is added to the steel (excluding Mn), the copper content is preferably 0.01 ~ 0.2 %.

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According to the present invention, the contents and the combination of Mn, Cu and S are controlled so as to create fine precipitates, and these are varied according to the added amount of Mn and/or Cu.

For the MnS-precipitated steel, the combination of Mn and S preferably satisfies the relationship: $0.58 \cdot \text{Mn}/\text{S} \leq 10$ (where Mn and S are denoted in terms of wt%). Mn combines with S to create the MnS precipitates. The MnS precipitates can be varied in a precipitated state according to the added amount of Mn and S, and thereby influence the bake hardenability, the yield strength, and the in-plane anisotropy index of the steel sheet. A value of $0.58 \cdot \text{Mn}/\text{S}$ greater than 10 creates coarse MnS precipitates, resulting in deterioration of the bake hardenability and the in-plane anisotropy index.

For the CuS-precipitated steel, the combination of Cu and S preferably satisfies the relationship: $1 \leq 0.5 \cdot \text{Cu}/\text{S} \leq 10$ (where Cu and S are denoted in terms of wt%). Cu combines with S to create the CuS precipitates, which can be varied in a precipitated state according to the added amount of Cu and S, and thereby influence the bake hardenability, the plasticity-anisotropy index, and the in-plane anisotropy index. A value of $0.5 \cdot \text{Cu}/\text{S}$ of 1 or more enables effective CuS precipitates to be created, and a value of $0.5 \cdot \text{Cu}/\text{S}$ greater than 10 creates coarse CuS precipitates, resulting in deterioration of the bake hardenability, the plasticity-anisotropy index, and of the in-plane anisotropy index. In order to stably secure the CuS precipitates having a size of $0.1 \mu\text{m}$ or less, the value of $0.5 \cdot \text{Cu}/\text{S}$ is preferably 1 ~ 3.

When Mn is added to the steel sheet together with Cu, the total content of Mn and Cu is preferably 0.3 % or less. This is attributed to the fact that a total Mn and Cu content exceeding 0.3 % is likely to create coarse precipitates, thereby detracting from the bake hardenability. The value of $0.5 \cdot (\text{Mn} + \text{Cu})/\text{S}$ (where Mn, Cu, and S are denoted in terms of wt%) is preferably 2 ~ 20. Mn and Cu combine with S to create the MnS, CuS, and (Mn, Cu)S precipitates, which can be varied in a precipitated state

according to the added amount of Mn, Cu, and S, and influence the bake hardenability, the plasticity-anisotropy index, and the in-plane anisotropy index. A value of $0.5 \cdot (\text{Mn} + \text{Cu}) / \text{S}$ of 2 or more enables effective precipitates to be created, and a value of $0.5 \cdot (\text{Mn} + \text{Cu}) / \text{S}$ exceeding 20 creates coarse precipitates, resulting in deterioration of the bake hardenability, the plasticity-anisotropy index, and the in-plane anisotropy index. According to the present invention, with the value of $0.5 \cdot (\text{Mn} + \text{Cu}) / \text{S}$ in the range of 2 ~ 20, the average size of the precipitates is reduced to 0.2 μm or less. In this case, it is desirable that 2×10^6 or more precipitates per unit area (number/ mm^2) are distributed in the grain. Below or above 7 as the value of $0.5 \cdot (\text{Mn} + \text{Cu}) / \text{S}$, the kinds and the number of the precipitates are remarkably varied. Specifically, when the value of $0.5 \cdot (\text{Mn} + \text{Cu}) / \text{S}$ is 7 or less, lots of very fine MnS and CuS separate precipitates are uniformly distributed rather than the (Mn, Cu)S complex precipitates. Meanwhile, when the value of $0.5 \cdot (\text{Mn} + \text{Cu}) / \text{S}$ is more than 7, regardless of a low difference between the sizes of the precipitates, the number of precipitates distributed in the grain is decreased because of an increase in amount of the (Mn, Cu)S complex precipitates. In the present invention, an increase in the number of precipitates can enhance the bake hardenability, the in-plane anisotropy index, the secondary work embrittlement resistance, and the like. For this purpose, it is preferred that 2×10^8 or more precipitates per unit area (number/ mm^2) are distributed in the grain and grain boundary. In the present invention, even in the case where the values of $0.5 \cdot (\text{Mn} + \text{Cu}) / \text{S}$ are the same, a smaller added amount of Mn and Cu can reduce the number of distributed precipitates. If the content of Mn and Cu is increased, the precipitates become coarse, leading to a reduction in the number of distributed precipitates.

According to the present invention, the MnS, CuS, and (Mn, Cu)S precipitates preferably have an average size of 0.2 μm or less. The MnS, CuS, and (Mn, Cu)S precipitates can have different appropriate sizes according to an added amount of Mn and Cu. Most preferably, the precipitates have a size of 0.2 μm or less for the MnS precipitates, a size of 0.1 μm or less for the CuS precipitates, and a size of 0.2 μm or less for the mixture of MnS, CuS, and (Mn, Cu)S precipitates. According to the present invention, if the MnS, CuS, and (Mn, Cu)S precipitates have an average size greater than a preferred size thereof, the bake hardenability is particularly deteriorated,

as well as deteriorating the plasticity-anisotropy index and the in-plane anisotropy index. As the size of the precipitates is reduced, it is preferred in terms of the bake hardenability.

5 Meanwhile, according to the present invention, for application to a high strength steel sheet of a 340 MPa-grade or more, at least one of the solid solution strengthening elements, that is, at least one of P, Si, and Cr may be added to the steel sheet. As the effects obtained by adding phosphorus were previously described, the description thereof will be omitted.

10

Silicon (Si): 0.1 ~ 0.8 %

Si is an alloying element, which can increase the solid solution strengthening effect while allowing a slight reduction in ductility, thus ensuring high strength of the steel sheet in which the precipitates are controlled according to the present invention.

15 The silicon content of 0.1 % or more can ensure the strength of the steel sheet, but a silicon content exceeding 0.8 % can cause a reduction in the ductility thereof.

Chrome (Cr): 0.2 ~ 1.2 %

20 Cr is an alloying element, which can increase solid solution strengthening effects while enhancing aging resistance at room temperature, thus ensuring the high strength of the steel sheet while reducing the in-plane anisotropy index of the steel sheet in which the precipitates are controlled according to the present invention. The chrome content of 0.2 % or more can ensure the strength of the steel sheet, but a chrome content of more than 1.2 % can cause the reduction in the ductility thereof.

25

Molybdenum (Mo) may be added to the cold rolled steel sheet of the present invention.

Molybdenum: 0.01 ~ 0.2 %.

30 Mo is an alloying element, which can increase the plasticity-anisotropy index of the steel sheet. The molybdenum content of 0.01 % or more can increase the plasticity-anisotropy index, but a molybdenum content exceeding 0.2 % can cause hot shortness without any additional improvement in the plasticity-anisotropy index.

[Method of manufacturing cold rolled steel sheet]

The present invention is characterized in that steel sheets satisfying the above-described compositions may be processed to have a finely reduced average size of precipitates through hot rolling and cold rolling. The average size of the precipitates is influenced by the contents and composition of Mn, Cu, and S, and the manufacturing process, and in particular, is directly influenced by a cooling rate after hot rolling.

[Hot rolling conditions]

According to the present invention, the steel satisfying the above-described compositions is reheated, followed by hot rolling. The reheating temperature is preferably 1,100 °C or more. This is attributed to the fact that a reheating temperature lower than 1,100 °C causes coarse precipitates to be created during continuous casting and to remain in an incompletely dissolved state, whereby the coarse precipitates remain even after the hot rolling.

Preferably, the hot rolling is performed under the condition that finish rolling is performed at an Ar_3 transformation temperature or more. If the finish rolling is performed below the Ar_3 transformation temperature, rolled grains are created, and remarkably lower the ductility as well as the formability of the steel sheet.

The cooling rate is preferably 200 °C/min or more after the hot rolling. More specifically, there is a slight difference between the cooling rates of (1) MnS-precipitated steel, (2) CuS-precipitated steel, and (3) MnCu-precipitated steel.

First, in the case of the MnS-precipitated steel, the cooling rate is preferably 200 °C/min or more. Even when the composition of Mn and S satisfies the relationship: $0.58 \cdot \text{Mn}/\text{S} \leq 10$ according to the present invention, a cooling rate lower than 200 °C/min can create coarse MnS precipitates having a size greater than 0.2 μm . That is, as the cooling rate is increased, a number of nuclei are created, so that the MnS precipitates become finer. When the composition of Mn and S has the relationship: $0.58 \cdot \text{Mn}/\text{S} > 10$, the number of coarse precipitates in an incompletely dissolved state during the reheating process is increased, so that, even if the cooling rate is increased, the number of nuclei is not increased, and thus the MnS precipitates do not become any finer (Fig. 2b, 0.0038 % C; 0.43 % Mn; 0.011 % P; 0.009 % S; 0.035 % Al; and 0.0043 % N).

Referring to Figs. 2a and 2b, since an increase of the cooling rate leads to creation of finer MnS precipitates, it is not necessary to provide an upper limit of the cooling rate. However, even when the cooling rate is 1,000 °C/min or more, the MnS precipitates are not further reduced in size, and thus the cooling rate is more preferably
5 in the range of 200 ~ 1,000 °C/min.

Second, in the case of the CuS-precipitated steel, the cooling rate is preferably 300 °C/min or more after the hot rolling. Even when the composition of Cu and S satisfies the relationship: $0.5 \cdot \text{Cu/S} \leq 10$ according to the present invention, a cooling rate lower than 300 °C/min creates coarse CuS precipitates having a size greater than
10 0.1 μm. That is, as the cooling rate is increased, a number of nuclei are created, so that the CuS precipitates become finer. When the composition of Cu and S has the relationship: $0.5 \cdot \text{Cu/S} > 10$, the number of coarse precipitates in an incompletely dissolved state during the reheating process is increased, so that an increase of the cooling rate does not result in an increase of the number of nuclei, and thus the CuS
15 precipitates do not become any finer (Fig. 3c, 0.0039 % C; 0.01 % P; 0.005 % S; 0.03 % Al; 0.0015 % N; and 0.28 % Cu).

Referring to Figs. 3a to 3c, since an increase of the cooling rate leads to creation of finer CuS precipitates, it is not necessary to provide an upper limit of the cooling rate. However, even when the cooling rate is 1,000 °C/min or more, the CuS
20 precipitates are not further reduced in size, and thus the cooling rate is more preferably in the range of 300 ~ 1,000 °C/min. Figs. 3a and 3b (0.0043 % C; 0.01 % P; 0.005 % S; 0.03 % Al; 0.0024 % N; and 0.081 % Cu) show the cases of $0.5 \cdot \text{Cu/S} \leq 3$, and of $0.5 \cdot \text{Cu/S} > 3$, respectively. As shown in the drawings, it can be seen that, when the value of $0.5 \cdot \text{Cu/S}$ is 3 or less, the CuS precipitates having a size of 0.1 μm or less can
25 be more stably obtained.

Last, in the case of the MnCu-precipitated steel, the cooling rate is preferably 300 °C/min or more. Even when the composition of Mn, Cu and S satisfies the relationship: $2 \leq 0.5 \cdot (\text{Mn} + \text{Cu})/\text{S} \leq 20$ according to the present invention, a cooling rate lower than 300 °C/min creates coarse precipitates having an average size greater than
30 0.2 μm. That is, as the cooling rate is increased, a number of nuclei are created, so that the precipitates become finer. When the composition of Mn, Cu and S has the relationship: $0.5 \cdot (\text{Mn} + \text{Cu})/\text{S} > 20$, the coarse precipitates in an incompletely dissolved

state during the reheating process are increased, so that, even if the cooling rate is increased, the number of nuclei is not increased, and thus the precipitates do not become any finer (Fig. 4b, 0.0039 % C; 0.4 % Mn; 0.01 % P; 0.01 % S; 0.05 % Al; 0.0016 % N; and 0.15 % Cu).

5 Referring to Figs. 4a and 4b, since an increase of the cooling rate leads to creation of finer precipitates, it is not necessary to provide an upper limit of the cooling rate. However, even when the cooling rate is 1,000 °C/min or more, the precipitates are not further reduced in size, and thus the cooling rate is more preferably in the range of 300 ~ 1,000 °C/min.

10

[Winding conditions]

After the hot rolling described above, the winding process is preferably performed at a temperature of 700 °C or less. When the winding process is performed at a temperature higher than 700 °C, the precipitates are grown too coarsely,
15 thereby reducing the bake hardenability of the steel.

[Cold rolling conditions]

The steel is cold rolled to a desired thickness, preferably at a reduction rate of 50 ~ 90 %. Since a reduction rate lower than 50 % leads to creation of a small amount
20 of nuclei upon recrystallization annealing, the crystal grains are grown excessively upon annealing, thereby coarsening of the crystal grains recrystallized through annealing, which results in reduction of the strength and formability. A cold reduction rate more than 90 % leads to enhanced formability, while creating an excessive number of nuclei, so that the crystal grains recrystallized through annealing become excessively
25 fine, thereby reducing the ductility of the steel.

[Continuous annealing]

The continuous annealing temperature plays an important role in determining the mechanical properties of the products. According to the present invention, the
30 continuous annealing is preferably performed at a temperature of 500 ~ 900 °C. Continuous annealing at a temperature lower than 500 °C creates excessively fine recrystallized crystal grains, so that a desired ductility cannot be ensured. Continuous

annealing at a temperature higher than 900 °C creates coarse recrystallized crystal grains, so that the strength of the steel is reduced. Holding time upon the continuous annealing is provided so as to complete the recrystallization of the steel, and the recrystallization of the steel can be completed for about 10 seconds or more upon the
 5 continuous annealing.

The present invention will be described in detail with reference to examples as follows.

In the following description of the examples, cold rolled steel sheets were
 10 machined to standard samples according to ASTM standards (ASTM E-8 Standard), and the mechanical properties thereof were measured. The mechanical properties were measured by use of a tensile strength tester (available from INSTRON Company, Model No. 6025). Among the mechanical strengths, the post-bake yield strength was measured, after the samples were subjected to 2 % strain, followed by the heat
 15 treatment at 120 °C for twenty minutes. The plasticity-anisotropy index (r_m value), and the in-plane anisotropy index (Δr value) were obtained according to the following equations (1) and (2), respectively:

$$r_m = (r_0 + 2 r_{45} + r_{90})/4 \quad \text{----} \quad (1)$$

$$\Delta r = (r_0 - 2 r_{45} + r_{90})/2 \quad \text{----} \quad (2)$$

20 Meanwhile, the average size and number of precipitates were obtained after measuring the size and number of all precipitates existing in the matrix.

[Example 1-1] MnS-precipitated steel

In order to provide MnS-precipitated steel according to the present invention,
 25 after steel slabs shown in Table 1 were reheated to a temperature of 1,200 °C, followed by finish rolling the steel slabs in order to provide hot rolled steel sheets, the hot rolled steel sheets were cooled at a cooling rate of 200 °C/min, and then coiled at 650 °C. Then, the hot rolled steel sheets were cold rolled at a reduction rate of 75 %, followed by continuous annealing the cold rolled steel sheets. The finish rolling was performed
 30 at 910 °C, which is above the A_{r3} transformation temperature, and the continuous annealing was performed by heating the steel sheets to 750 °C at a velocity of 10 °C/second for 40 seconds. Exceptionally, for sample A8 in Table 1, after

reheating to a temperature of 1,050 °C, followed by finish rolling, it was cooled at a cooling rate of 50 °C/minute, and was then coiled at 750 °C.

Table 1

5

Sample No.	Component (wt%)							
	C	Mn	P	S	Al	N	Mo	R-1
	0.003-0.005	0.05-0.2	≤0.015	0.005-0.03	0.01-0.1	≤0.004	0.01-0.2	≤10
A1	0.0035	0.1	0.01	0.009	0.04	0.0021	-	6.44
A2	0.0041	0.10	0.009	0.012	0.05	0.0030	-	4.83
A3	0.0038	0.08	0.011	0.012	0.035	0.0023	-	3.87
A4	0.0044	0.1	0.01	0.006	0.02	0.0032	-	9.67
A5	0.0022	0.1	0.009	0.011	0.04	0.0038	-	5.27
A6	0.0039	0.43	0.01	0.008	0.05	0.0038	-	31.2
A7	0.0067	0.1	0.008	0.01	0.04	0.0028	-	5.8
A8	0.0024	0.4	0.07	0.01	0.04	0.0016	Ti 0.02	11.6
A9	0.0042	0.11	0.012	0.01	0.032	0.0018	0.02	6.38
A10	0.0038	0.1	0.01	0.008	0.035	0.0025	0.16	7.25
A11	0.0045	0.08	0.011	0.011	0.04	0.0011	0.064	4.22
A12	0.0044	0.08	0.01	0.01	0.025	0.003	0.092	4.64
A13	0.0046	0.09	0.012	0.012	0.04	0.0015	0.27	4.35

Note: R-1 = 0.58*Mn/S

Table 2

Sample No.	Mechanical properties							AS (μm)	Remarks
	YS (Mpa)	TS (MPa)	El (%)	r-value (r _m)	Δr-value (Δr)	PBYS (Mpa)	DBTT (°C)		
A1	221	310	49	1.83	0.41	288	- 70	0.13	IS
A2	241	315	47	1.75	0.36	292	- 70	0.14	IS
A3	233	312	47	1.73	0.38	282	- 70	0.12	IS
A4	245	328	45	1.69	0.31	301	- 70	0.1	IS
A5	209	299	51	1.88	0.42	232	- 70	0.11	CS

A6	211	290	52	1.82	0.38	235	- 70	0.59	CS
A7	251	329	42	1.53	0.29	298	- 70	0.12	CS
A8	182	292	48	1.83	0.58	215	- 10	0.21	CVS
A9	225	312	49	2.13	0.35	273	- 80	0.12	IS
A10	229	310	48	2.21	0.28	270	- 80	0.13	IS
A11	235	318	47	2.09	0.31	281	- 80	0.11	IS
A12	236	320	46	2.01	0.33	304	- 80	0.1	IS
A13	232	328	46	1.73	0.35	305	-80	0.12	CS

Note: YS = Yield strength, TS = Tensile strength, El = Elongation, r-value: Plasticity-anisotropy index, Δr -value: In-plane anisotropy index, PBYS = Post-bake yield strength, DBTT = ductility-brittleness transition temperature for investigating secondary work embrittlement, AS = Average size of precipitates, IS = Steel of the invention, CS = Comparative steel, CVS = Conventional steel

5

As shown in Table 2, samples A1 ~ A4 have excellent yield strength, elongation ratio, and yield strength-ductility balance as well as the bake hardenability. Additionally, these samples have a high plasticity-anisotropy index and a low in-plane anisotropy index, thereby providing excellent formability.

10

On the contrary, due to its low carbon content, sample A5 provides low post-bake yield strength. Due to its large size of the precipitates, sample A6 also has low post-bake yield strength. Due to its high carbon content, sample A7 has low elongation ratio and plasticity anisotropy index, thereby providing a high possibility of fracture during the forming process. Due to its low post-bake yield strength and high secondary work embrittlement temperature, sample A8, which is a conventional IF steel sheet, provides a high possibility of fracture upon impact.

15

Samples A9 to A12 have excellent formability together with the bake hardenability. On the contrary, due to its high amount of Mo added, sample A13 has poor formability.

20

[Example 1-2] High strength MnS-precipitated steel with solid solution strengthening

25

In order to provide high strength MnS-precipitated steel with solid solution strengthening according to the present invention, after steel slabs shown in Table 3 were reheated to a temperature of 1,200 °C, followed by finish rolling the steel slabs in order to provide hot rolled steel sheets, the hot rolled steel sheets were cooled at a cooling rate of 200 °C/min, and then coiled at 650 °C. Then, the hot rolled steel sheets were cold rolled at a reduction rate of 75 %, followed by continuous annealing

the cold rolled steel sheets. The finish rolling was performed at 910 °C, which is above the Ar₃ transformation temperature, and the continuous annealing was performed by heating the steel sheets to 750 °C at a velocity of 10 °C/second for 40 seconds.

5 Table 3

Sample No.	Component (wt%)									
	C	Mn	P	Si	Cr	S	Al	N	Mo	R-1
	0.003-0.005	0.05-0.2	≤0.2	0.1-0.8	0.2-1.2	0.005-0.03	0.01-0.1	≤0.004	0.01-0.2	≤10
B1	0.0035	0.08	0.052	-	-	0.006	0.04	0.0015	-	7.73
B2	0.0042	0.10	0.102	-	-	0.010	0.05	0.0026	-	5.8
B3	0.0039	0.08	0.151	-	-	0.012	0.035	0.0018	-	3.87
B4	0.0018	0.52	0.052	-	-	0.011	0.03	0.0039	-	29
B5	0.0058	0.44	0.11	-	-	0.011	0.05	0.0025	-	21.1
B6	0.0038	0.38	0.15	-	-	0.008	0.05	0.0028	-	31.2
B7	0.0039	0.09	0.009	0.24	-	0.006	0.05	0.0022	-	8.7
B8	0.0042	0.09	0.013	0.43	-	0.012	0.03	0.0026	-	4.35
B9	0.0035	0.1	0.011	0.62	-	0.009	0.035	0.0025	-	6.4
B10	0.0022	0.4	0.009	0.25	-	0.009	0.03	0.0042	-	25.8
B11	0.0077	0.42	0.01	0.44	-	0.011	0.04	0.0042	-	21.1
B12	0.0042	0.4	0.01	0.62	-	0.009	0.05	0.0039	-	25.8
B13	0.0044	0.1	0.01	-	0.35	0.007	0.04	0.0024	-	8.29
B14	0.0032	0.09	0.01	-	0.65	0.012	0.04	0.0032	-	4.35
B15	0.0038	0.11	0.012	-	0.82	0.017	0.05	0.0018	-	3.75
B16	0.0025	0.4	0.011	-	0.32	0.009	0.03	0.0017	-	25.8
B17	0.0059	0.43	0.01	-	0.62	0.012	0.05	0.0024	-	20.8
B18	0.0042	0.4	0.01	-	0.82	0.01	0.04	0.0019	-	31.2
B19	0.0044	0.1	0.05	-	-	0.008	0.034	0.0018	0.025	7.25
B20	0.0046	0.1	0.01	0.25	-	0.008	0.035	0.0032	0.028	7.25

B21	0.0034	0.11	0.011	-	0.33	0.009	0.034	0.0012	0.019	7.09
B22	0.0041	0.11	0.045	0.21	0.35	0.01	0.03	0.0022	0.08	6.38
B23	0.0038	0.1	0.048	0.18	0.35	0.012	0.042	0.0035	0.06	4.83
B24	0.0043	0.1	0.012	0.2	0.32	0.015	0.029	0.0018	0.04	3.87

Note: $R-I = 0.58 \cdot Mn/S$

Table 4

Sample No.	Mechanical properties							AS (μm)	Remarks
	YS (MPa)	TS (MPa)	El (%)	r-value (r_m)	Δr -value (Δr)	PBYS (MPa)	DBTT ($^{\circ}C$)		
B1	252	362	43	1.65	0.25	304	- 70	0.13	IS
B2	305	411	36	1.52	0.29	346	- 50	0.12	IS
B3	377	460	32	1.46	0.27	414	- 40	0.09	IS
B4	235	342	44	1.71	0.44	258	- 60	0.59	CS
B5	302	409	33	1.39	0.22	359	- 60	0.73	CS
B6	352	450	32	1.40	0.46	381	- 40	0.59	CS
B7	250	360	45	1.64	0.25	312	- 80	0.09	IS
B8	315	421	40	1.52	0.22	348	- 60	0.11	IS
B9	366	460	35	1.46	0.29	414	- 50	0.11	IS
B10	238	342	47	1.73	0.62	255	- 70	0.52	CS
B11	324	430	31	1.40	0.28	358	- 60	0.45	CS
B12	340	440	35	1.42	0.42	360	- 40	0.62	CS
B13	239	360	44	1.62	0.20	293	- 80	0.09	IS
B14	306	420	38	1.44	0.22	359	- 60	0.10	IS
B15	350	462	33	1.40	0.21	428	- 50	0.09	IS
B16	230	345	46	1.68	0.42	250	- 70	0.42	CS
B17	319	429	32	1.32	0.22	368	- 60	0.35	CS
B18	342	459	28	1.25	0.13	382	- 40	0.42	CS
B19	259	361	44	1.95	0.31	321	- 80	0.12	IS
B20	255	355	46	1.98	0.32	302	- 80	0.1	IS
B21	235	359	46	1.95	0.29	295	- 80	0.09	IS
B22	351	474	36	1.59	0.17	406	- 60	0.1	IS
B23	335	462	35	1.55	0.15	390	- 60	0.11	IS
B24	328	419	39	1.67	0.19	358	- 70	0.09	IS

Note: YS = Yield strength, TS = Tensile strength, El = Elongation, r-value: Plasticity-anisotropy index, Δr -value: In-plane

anisotropy index, PBYS = Post-bake yield strength, DBTT = Ductility-brittleness transition temperature for investigating secondary work embrittlement, AS = Average size of precipitates, IS = Steel of the invention, CS = Comparative steel

[Example 1-3] MnS-precipitated steel with AlN precipitation strengthening

5 In order to provide MnS-precipitated steel with AlN precipitation strengthening according to the present invention, after steel slabs shown in Table 5 were reheated to a temperature of 1,200 °C, followed by finish rolling the steel slabs in order to provide hot rolled steel sheets, the hot rolled steel sheets were cooled at a cooling rate of 200 °C/min, and then coiled at 650 °C. Then, the hot rolled steel
10 sheets were cold rolled at a reduction rate of 75 %, followed by continuous annealing the cold rolled steel sheets. The finish rolling was performed at 910 °C, which is above the Ar₃ transformation temperature, and the continuous annealing was performed by heating the steel sheets to 750 °C at a velocity of 10 °C/second for 40 seconds.

15 Table 5

Sample No.	Component (wt%)								
	C	Mn	P	S	Al	N	Mo	R-1	R-2
	0.003-0.005	0.05-0.2	0.03-0.06	0.005-0.03	0.01-0.1	0.005-0.02	0.01-0.2	≤ 10	1-5
C1	0.0045	0.1	0.035	0.01	0.04	0.0135	-	5.8	1.54
C2	0.0038	0.11	0.044	0.007	0.055	0.0079	-	9.11	3.63
C3	0.0042	0.08	0.053	0.009	0.055	0.0065	-	3.87	4.4
C4	0.0018	0.10	0.042	0.01	0.04	0.0014	-	5.8	14.9
C5	0.0075	0.09	0.04	0.011	0.008	0.0067	-	6.53	0.46
C6	0.0035	0.4	0.04	0.009	0.04	0.0083	-	25.8	2.51
C7	0.0047	0.11	0.044	0.009	0.044	0.011	0.03	7.09	2.08
C8	0.0037	0.1	0.042	0.01	0.05	0.012	0.064	5.8	2.17
C9	0.0044	0.09	0.04	0.01	0.042	0.01	0.15	5.22	1.05
C10	0.004	0.11	0.04	0.009	0.04	0.01	0.27	7.09	2.08

Note: R-1 = 0.58*Mn/S, R-2 = 0.52*Al/N

Table 6

Sample No.	Mechanical properties							AS (μm)	Remarks
	YS (MPa)	TS (MPa)	El (%)	r-value (r_m)	Δr -value (Δr)	PBYS (MPa)	DBTT ($^{\circ}\text{C}$)		
C1	242	358	44	1.71	0.31	283	- 70	0.07	IS
C2	224	355	43	1.75	0.38	280	- 70	0.09	IS
C3	239	360	40	1.68	0.29	302	- 70	0.11	IS
C4	210	330	46	1.78	0.32	269	- 70	0.11	CS
C5	228	352	37	1.52	0.25	295	- 70	0.12	CS
C6	228	360	40	1.65	0.54	280	- 70	0.41	CS
C7	246	362	45	2.09	0.34	298	- 80	0.08	IS
C8	220	350	46	2.18	0.42	287	- 80	0.07	IS
C9	230	357	44	2.00	0.32	276	- 80	0.11	IS
C10	239	362	43	1.79	0.27	300	- 80	0.1	CS

Note: YS = Yield strength, TS = Tensile strength, El = Elongation, r-value: Plasticity-anisotropy index, Δr -value: In-plane anisotropy index, PBYS = Post-bake yield strength, DBTT = ductility-brittleness transition temperature for investigating secondary

5 work embrittlement, AS = Average size of precipitates, IS = Steel of the invention, CS = Comparative steel

[Example 2-1] CuS-precipitated steel

In order to provide CuS-precipitated steel according to the present invention, after steel slabs shown in Table 7 were reheated to a temperature of 1,200 $^{\circ}\text{C}$, followed
 10 by finish rolling the steel slabs in order to provide hot rolled steel sheets, the hot rolled steel sheets were cooled at a cooling rate of 400 $^{\circ}\text{C}/\text{min}$, and then coiled at 650 $^{\circ}\text{C}$. Then, the hot rolled steel sheets were cold rolled at a reduction rate of 75 %, followed by continuous annealing the cold rolled steel sheets. The finish rolling was performed at 910 $^{\circ}\text{C}$, which is above the Ar_3 transformation temperature, and the continuous
 15 annealing was performed by heating the steel sheets to 750 $^{\circ}\text{C}$ at a velocity of 10 $^{\circ}\text{C}/\text{second}$ for 40 seconds. Exceptionally, for sample D7 in Table 7, after reheating to a temperature of 1,050 $^{\circ}\text{C}$, followed by finish rolling, it was cooled at a cooling rate of 400 $^{\circ}\text{C}/\text{minute}$, and was then coiled at 650 $^{\circ}\text{C}$. For samples D8 to D11 in Table 7, after reheating to a temperature of 1,200 $^{\circ}\text{C}$, followed by finish rolling,
 20 it was cooled at a cooling rate of 450 $^{\circ}\text{C}/\text{minute}$, and was then coiled at 650 $^{\circ}\text{C}$.

Table 7

Sample No.	Component (wt%)							
	C	P	S	Al	N	Cu	Mo	R-3
	0.003-0.005	≤ 0.015	0.003-0.025	0.01-0.1	≤ 0.004	0.01-0.2	0.01-0.2	1-10
D1	0.0038	0.01	0.01	0.04	0.0025	0.035	-	1.75
D2	0.0045	0.009	0.008	0.04	0.0026	0.045	-	2.81
D3	0.0035	0.011	0.006	0.03	0.0012	0.06	-	5.0
D4	0.0042	0.009	0.005	0.04	0.0027	0.083	-	8.3
D5	0.0016	0.011	0.009	0.05	0.0038	0.05	-	2.78
D6	0.0037	0.009	0.008	0.04	0.0015	0.25	-	15.6
D7	0.0078	0.010	0.012	0.04	0.0024	0.064	-	2.67
D8	0.0035	0.011	0.009	0.024	0.0035	0.038	0.018	2.11
D9	0.0043	0.009	0.011	0.043	0.0026	0.04	0.083	1.82
D10	0.0039	0.01	0.01	0.038	0.0042	0.062	0.17	3.1
D11	0.004	0.012	0.011	0.028	0.0032	0.053	0.25	2.41

Note: R-3 = 0.5*Cu/S

Table 8

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Sample No.	Mechanical properties							AS (μm)	Remarks
	YS (Mpa)	TS (MPa)	El (%)	r-value (r_m)	Δr -value (Δr)	PBYS (MPa)	DBTT ($^{\circ}\text{C}$)		
D1	219	310	49	1.88	0.41	265	- 70	0.08	IS
D2	224	325	47	1.83	0.36	275	- 70	0.08	IS
D3	225	330	45	1.79	0.38	289	- 70	0.07	IS
D4	240	335	45	1.75	0.30	311	- 70	0.09	IS
D5	205	290	50	1.88	0.46	235	- 70	0.09	CS
D6	216	299	49	1.80	0.38	240	- 70	0.48	CS
D7	256	339	40	1.53	0.29	320	- 70	0.08	CS
D8	214	310	48	2.10	2.9	260	- 70	0.07	IS
D9	228	320	47	2.01	3.1	271	- 70	0.07	IS
D10	220	325	46	1.99	2.7	279	- 70	0.08	IS
D11	219	319	48	1.71	2.6	285	- 70	0.1	CS

Note: YS = Yield strength, TS = Tensile strength, El = Elongation, r-value: Plasticity-anisotropy index, Δr -value: In-plane

anisotropy index, PBYS = Post-bake yield strength, DBTT = ductility-brittleness transition temperature for investigating secondary work embrittlement, AS = Average size of precipitates, IS = Steel of the invention, CS = Comparative steel

5 [Example 2-2] High strength CuS-precipitated steel with solid solution strengthening

In order to provide high strength CuS-precipitated steel with solid solution strengthening according to the present invention, after steel slabs shown in Table 9 were reheated to a temperature of 1,200 °C, followed by finish rolling the steel slabs in order to provide hot rolled steel sheets, the hot rolled steel sheets were cooled at a cooling rate of 400 °C/min, and then coiled at 650 °C. Then, the hot rolled steel sheets were cold rolled at a reduction rate of 75 %, followed by continuous annealing the cold rolled steel sheets. The finish rolling was performed at 910 °C, which is above the Ar₃ transformation temperature, and the continuous annealing was performed by heating the steel sheets to 750 °C at a velocity of 10 °C/second for 40 seconds.

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Table 9

Sample No.	Component (wt%)									
	C	P	Si	Cr	S	Al	N	Cu	Mo	R-3
	0.003-0.005	≤0.2	0.1-0.8	0.2-1.2	0.003-0.025	0.01-0.1	≤0.004	0.01-0.2	0.01-0.2	1-10
E1	0.0038	0.050	-	-	0.006	0.04	0.0025	0.095	-	7.92
E2	0.0046	0.11	-	-	0.008	0.03	0.0026	0.06	-	3.75
E3	0.0033	0.148	-	-	0.01	0.04	0.0018	0.038	-	1.9
E4	0.0018	0.050	-	-	0.011	0.04	0.0024	0.054	-	2.45
E5	0.0065	0.115	-	-	0.009	0.03	0.0025	0.082	-	4.56
E6	0.0038	0.15	-	-	0.006	0.05	0.0028	0.25	-	20.8
E7	0.0039	0.01	0.25	-	0.006	0.04	0.0026	0.1	-	8.25
E8	0.0042	0.011	0.45	-	0.008	0.05	0.0016	0.086	-	5.35
E9	0.0035	0.015	0.65	-	0.012	0.05	0.0028	0.051	-	2.14
E10	0.0018	0.009	0.25	-	0.009	0.03	0.0042	0.077	-	4.28
E11	0.0077	0.011	0.42	-	0.011	0.04	0.0042	0.046	-	2.09

E12	0.0042	0.01	0.62	-	0.007	0.05	0.0039	0.252	-	18
E13	0.0035	0.009	-	0.34	0.005	0.05	0.0014	0.08	-	8.00
E14	0.0038	0.011	-	0.62	0.01	0.04	0.0022	0.09	-	4.5
E15	0.0045	0.01	-	0.83	0.018	0.04	0.0028	0.08	-	2.22
E16	0.0016	0.01	-	0.34	0.011	0.03	0.0017	0.08	-	3.64
E17	0.0072	0.009	-	0.59	0.008	0.04	0.0026	0.12	-	7.5
E18	0.0035	0.012	-	0.80	0.005	0.05	0.0013	0.26	-	26
E19	0.0045	0.054	-	-	0.008	0.024	0.0022	0.06	0.02	3.75
E20	0.0036	0.011	0.27	-	0.008	0.034	0.0028	0.06	0.018	3.75
E21	0.0032	0.01	-	0.32	0.011	0.035	0.0025	0.054	0.022	1.23

Note: R-3 = 0.5*Cu/S

Table 10

Sample No.	Mechanical properties							AS (μm)	Remarks
	YS (MPa)	TS (MPa)	El (%)	r-value (rm)	Δ r-value (Δ r)	PBYS (MPa)	DBTT ($^{\circ}\text{C}$)		
E1	248	360	43	1.70	0.28	310	- 70	0.09	IS
E2	308	405	37	1.58	0.28	340	- 50	0.08	IS
E3	367	465	33	1.46	0.26	410	- 40	0.07	IS
E4	230	340	45	1.70	0.44	262	- 60	0.49	CS
E5	322	415	30	1.35	0.20	365	- 60	0.73	CS
E6	345	445	32	1.40	0.46	380	- 40	0.59	CS
E7	250	350	47	1.69	0.23	318	- 80	0.09	IS
E8	325	415	42	1.55	0.22	368	- 60	0.08	IS
E9	356	465	35	1.44	0.21	424	- 50	0.08	IS
E10	238	352	47	1.69	0.58	250	- 70	0.45	CS
E11	325	425	32	1.40	0.28	358	- 60	0.45	CS
E12	345	445	32	1.42	0.42	370	- 40	0.62	CS
E13	242	365	43	1.69	0.21	302	- 80	0.09	IS
E14	310	425	38	1.46	0.23	365	- 60	0.08	IS
E15	352	454	36	1.45	0.21	408	- 70	0.07	IS
E16	230	345	46	1.68	0.32	265	- 70	0.07	CS
E17	315	413	28	1.32	0.22	365	- 60	0.09	CS

E18	348	461	27	1.24	0.13	372	- 60	0.42	CS
E19	254	365	43	2.08	2.8	322	- 70	0.09	IS
E20	247	348	48	1.95	0.28	295	- 80	0.09	IS
E21	240	358	45	1.93	0.27	298	- 80	0.07	IS

Note: YS = Yield strength, TS = Tensile strength, El = Elongation, r-value: Plasticity-anisotropy index, Δr -value: In-plane anisotropy index, PBYS = Post-bake yield strength, DBTT = ductility-brittleness transition temperature for investigating secondary work embrittlement, AS = Average size of precipitates, IS = Steel of the invention, CS = Comparative steel

5 [Example 2-3] CuS-precipitated steel with AlN precipitation strengthening

In order to provide CuS-precipitated steel with AlN precipitation strengthening according to the present invention, after steel slabs shown in Table 11 were reheated to a temperature of 1,200 °C, followed by finish rolling the steel slabs in order to provide hot rolled steel sheets, the hot rolled steel sheets were cooled at a cooling rate of 400 °C/min, and then coiled at 650 °C. Then, the hot rolled steel sheets were cold rolled at a reduction rate of 75 %, followed by continuous annealing the cold rolled steel sheets. The finish rolling was performed at 910 °C, which is above the A_{r3} transformation temperature, and the continuous annealing was performed by heating the steel sheets to 750 °C at a velocity of 10 °C/second for 40 seconds. Exceptionally, for samples F8 to F10 of Table 11, after reheating to a temperature of 1,200 °C, followed by finish rolling, these samples were cooled at a cooling rate of 550 °C/minute, and then coiled at 650 °C.

Table 11

Sample No.	Component (wt%)								
	C	P	S	Al	N	Cu	Mo	R-3	R-2
	0.003-0.005	0.03-0.06	0.003-0.025	0.01-0.1	0.005-0.02	0.01-0.2	0.01-0.2	1-10	1-5
F1	0.0042	0.041	0.005	0.045	0.0125	0.09	-	9	1.87
F2	0.0035	0.042	0.008	0.05	0.0072	0.052	-	3.25	3.61
F3	0.0045	0.043	0.014	0.04	0.0052	0.07	-	2.5	4
F4	0.0015	0.04	0.01	0.05	0.0014	0.08	-	4	18.6

F5	0.0073	0.037	0.008	0.01	0.0077	0.1	-	6.25	0.68
F6	0.0036	0.042	0.006	0.04	0.0083	0.155	-	12.9	2.51
F7	0.0037	0.044	0.011	0.055	0.012	0.09	0.018	4.09	2.38
F8	0.0043	0.04	0.009	0.045	0.0092	0.088	0.078	4.89	2.54
F9	0.0035	0.044	0.012	0.054	0.011	0.097	0.16	4.04	2.55
F10	0.0045	0.042	0.008	0.053	0.0084	0.082	0.25	5.13	3.28

Note: R-2 = 0.52*Al/N, R-3 = 0.5*Cu/S

5 Table 12

Sample No.	Mechanical properties							AS (μm)	Remarks
	YS (MPa)	TS (MPa)	El (%)	r-value (rm)	Δ r-value (Δ r)	PBYS (MPa)	DBTT ($^{\circ}\text{C}$)		
F1	240	353	45	1.70	0.32	296	- 70	0.06	IS
F2	232	350	44	1.72	0.28	291	- 70	0.07	IS
F3	245	362	46	1.80	0.27	323	- 70	0.05	CS
F4	216	340	46	1.78	0.35	260	- 70	0.08	CS
F5	243	360	35	1.49	0.25	308	- 70	0.07	CS
F6	238	355	43	1.69	0.44	253	- 70	0.41	CS
F7	235	348	46	1.94	0.24	296	- 70	0.06	IS
F8	237	355	44	1.93	0.22	302	- 70	0.08	IS
F9	237	360	46	1.97	0.26	312	- 70	0.06	IS
F10	231	346	46	1.70	0.27	300	- 70	0.07	CS

Note: YS = Yield strength, TS = Tensile strength, El = Elongation, r-value: Plasticity-anisotropy index, Δ r-value: In-plane anisotropy index, PBYS = Post-bake yield strength, DBTT = ductility-brittleness transition temperature for investigating secondary work embrittlement, AS = Average size of precipitates, IS = Steel of the invention, CS = Comparative steel

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[Example 3-1] MnCu-precipitated steel

In order to provide MnCu-precipitated steel according to the present invention, after steel slabs shown in Table 13 were reheated to a temperature of 1,200 $^{\circ}\text{C}$, followed by finish rolling the steel slabs in order to provide hot rolled steel sheets, the hot rolled steel sheets were cooled at a cooling rate of 600 $^{\circ}\text{C}/\text{min}$, and then coiled at 650 $^{\circ}\text{C}$. Then, the hot rolled steel sheets were cold rolled at a reduction rate of 75 %, followed by continuous annealing the cold rolled steel sheets. The finish rolling was

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performed at 910 °C, which is above the Ar₃ transformation temperature, and the continuous annealing was performed by heating the steel sheets to 750 °C at a velocity of 10 °C/second for 40 seconds.

5 Table 13

Sample No.	Component (wt%)									
	C	Mn	P	S	Al	N	Cu	Mo	R-4	R-5
	0.003-0.005	0.03-0.2	≤0.015	0.003-0.025	0.01-0.1	≤0.004	0.01-0.2	0.01-0.2	≤0.3	2-20
G1	0.0041	0.08	0.012	0.02	0.05	0.0013	0.03	-	0.11	2.75
G2	0.0038	0.1	0.009	0.012	0.04	0.0021	0.04	-	0.14	5.83
G3	0.0044	0.15	0.01	0.015	0.04	0.0024	0.05	-	0.2	6.67
G4	0.0042	0.17	0.008	0.009	0.04	0.0012	0.12	-	0.29	16.1
G5	0.0012	0.12	0.01	0.012	0.05	0.0014	0.05	-	0.17	7.08
G6	0.0064	0.15	0.009	0.01	0.04	0.0023	0.03	-	0.18	9.0
G7	0.0042	0.45	0.01	0.011	0.05	0.0013	0.18	-	0.63	28.6
G8	0.0035	0.11	0.011	0.02	0.045	0.0032	0.03	0.019	0.14	3.5
G9	0.0033	0.12	0.01	0.014	0.034	0.0019	0.046	0.082	0.17	5.93
G10	0.0043	0.12	0.014	0.009	0.027	0.0034	0.062	0.16	0.18	10.1
G11	0.0045	0.15	0.008	0.014	0.033	0.0032	0.085	0.25	0.24	8.39

Note: R-4 = Mn+Cu, R-5 = 0.5*(Mn+Cu)/S

Table 14

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Sample No.	Mechanical properties						AS (μm)	PN (number/mm ²)	Remarks
	YS (Mpa)	TS (MPa)	El (%)	r-value (rm)	Δr-value (Δr)	PBYS (MPa)			
G1	218	301	52	2.12	0.28	288	0.07	3.5X10 ⁹	IS
G2	208	295	54	2.25	0.31	275	0.08	7.4X10 ⁸	IS
G3	225	314	52	2.18	0.26	298	0.08	9.2X10 ⁸	IS
G4	195	292	50	2.10	0.26	275	0.09	9.5X10 ⁶	IS
G5	185	282	50	2.63	0.53	229	0.09	7.5X10 ⁷	CS
G6	205	321	42	1.73	0.33	292	0.12	2.5X10 ⁸	CS

G7	185	292	44	1.83	0.49	222	0.42	6.3X10 ⁴	CS
G8	212	304	52	2.42	0.39	278	0.08	4.4X10 ⁶	IS
G9	210	297	53	2.53	0.34	269	0.09	3.5X10 ⁶	IS
G10	221	310	51	2.52	0.38	288	0.07	8.2X10 ⁶	IS
G11	219	301	48	2.12	0.32	269	0.08	3.8X10 ⁶	CS

Note: YS = Yield strength, TS = Tensile strength, El = Elongation, r-value: Plasticity-anisotropy index, Δr -value: In-plane anisotropy index, PBYS = Post-bake yield strength, DBTT = ductility-brittleness transition temperature for investigating secondary work embrittlement, AS = Average size of precipitates, IS = Steel of the invention, CS = Comparative steel

5 [Example 3-2] High strength MnCu-precipitated steel with solid solution strengthening

In order to provide high strength MnCu-precipitated steel with solid solution strengthening according to the present invention, after steel slabs shown in Table 15 were reheated to a temperature of 1,200 °C, followed by finish rolling the steel slabs in order to provide hot rolled steel sheets, the hot rolled steel sheets were cooled at a cooling rate of 600 °C/min, and then coiled at 650 °C. Then, the hot rolled steel sheets were cold rolled at a reduction rate of 75 %, followed by continuous annealing the cold rolled steel sheets. The finish rolling was performed at 910 °C, which is above the Ar₃ transformation temperature, and the continuous annealing was performed by heating the steel sheets to 750 °C at a velocity of 10 °C/second for 40 seconds.

Table 15

Sample No.	Component (wt%)											
	C	Mn	P	Si	Cr	S	Al	N	Cu	Mo	R-4	R-5
	0.003-0.005	0.03-0.2	≤0.2	0.1-0.8	0.2-1.2	0.003-0.025	0.01-0.1	≤0.004	0.005-0.2	0.01-0.2	≤0.3	2-20
H1	0.0042	0.06	0.039	-	-	0.015	0.05	0.0025	0.02		0.08	2.67
H2	0.0045	0.1	0.041	-	-	0.018	0.04	0.0023	0.03	-	0.13	3.61
H3	0.0037	0.12	0.09	-	-	0.016	0.05	0.0032	0.04		0.16	5
H4	0.0045	0.18	0.14	-	-	0.011	0.04	0.0028	0.1	-	0.28	12.7
H5	0.0018	0.1	0.04	-	-	0.012	0.05	0.0024	0.1	-	0.2	8.33
H6	0.0075	0.15	0.1	-	-	0.012	0.03	0.0022	0.06	-	0.21	8.75
H7	0.0043	0.3	0.14	-	-	0.008	0.04	0.0015	0.15		0.45	28.1

H8	0.004	0.09	0.04	-	-	0.013	0.035	0.0029	0.028	0.017	0.12	4.54
H9	0.0044	0.11	0.094	-	-	0.012	0.026	0.0035	0.047	0.072	0.16	6.54
H10	0.0037	0.12	0.145	-	-	0.01	0.042	0.0018	0.088	0.16	0.21	10.4
H11	0.0045	0.08	0.043	-	-	0.009	0.037	0.0032	0.035	0.25	0.16	6.39
H12	0.0041	0.06	0.01	0.18	-	0.018	0.04	0.0019	0.04	-	0.1	5
H13	0.0036	0.1	0.009	0.17	-	0.015	0.05	0.0026	0.03	-	0.13	7.22
H14	0.0038	0.13	0.012	0.35	-	0.015	0.04	0.0032	0.03	-	0.16	6.67
H15	0.0045	0.22	0.012	0.55	-	0.01	0.04	0.0032	0.05	-	0.27	13.5
H16	0.0021	0.12	0.009	0.2	-	0.011	0.05	0.0024	0.12	-	0.24	10.9
H17	0.0064	0.12	0.01	0.34	-	0.012	0.04	0.0028	0.07	-	0.19	7.9
H18	0.0044	0.25	0.012	0.53	-	0.009	0.05	0.0022	0.18	-	0.43	23.9
H19	0.0039	0.11	0.012	0.21	-	0.014	0.034	0.0029	0.044	0.017	0.15	5.5
H20	0.0045	0.12	0.009	0.32	-	0.011	0.042	0.0042	0.038	0.075	0.16	7.18
H21	0.0036	0.14	0.012	0.62	-	0.009	0.033	0.0022	0.063	0.16	0.20	11.3
H22	0.0042	0.09	0.013	0.2	-	0.01	0.038	0.0033	0.053	0.25	0.14	7.15
H23	0.0044	0.07	0.009	-	0.25	0.017	0.04	0.0018	0.03	-	0.1	2.94
H24	0.0039	0.11	0.01	-	0.24	0.015	0.03	0.0022	0.03	-	0.14	4.67
H25	0.0042	0.15	0.011	-	0.55	0.015	0.04	0.0023	0.05	-	0.2	6.67
H26	0.0046	0.18	0.012	-	0.86	0.01	0.04	0.0016	0.04	-	0.22	11
H27	0.0022	0.15	0.01	-	0.24	0.015	0.04	0.0026	0.1	-	0.25	8.33
H28	0.0067	0.15	0.012	-	0.52	0.011	0.05	0.0025	0.11	-	0.26	11.8
H29	0.0043	0.2	0.009	-	0.88	0.008	0.03	0.0013	0.15	-	0.35	21.9
H30	0.0039	0.09	0.012	-	0.23	0.012	0.034	0.0028	0.043	0.019	0.13	5.54
H31	0.0045	0.12	0.01	-	0.58	0.013	0.042	0.0033	0.056	0.079	0.18	6.77
H32	0.0037	0.15	0.009	-	0.83	0.011	0.023	0.0021	0.063	0.17	0.21	9.68
H33	0.0042	0.1	0.011	-	0.22	0.011	0.035	0.0031	0.073	0.27	0.17	7.86

Note: R-4 = Mn+Cu, R-5 = 05*(Mn+Cu)/S

Table 16

Sample No.	Mechanical properties							AS (μm)	PN (number/ mm^2)	Remarks
	YS (Mpa)	TS (MPa)	El (%)	r-value (rm)	Δ r-value (Δ r)	PBYS (MPa)	DBTT ($^{\circ}\text{C}$)			
H1	265	360	49	1.98	0.25	346	- 70	0.05	5.5×10^8	IS
H2	258	358	50	1.92	0.28	345	-70	0.05	4.0×10^8	IS
H3	308	410	43	1.71	0.21	394	- 60	0.06	2.2×10^8	IS
H4	335	442	37	1.60	0.19	428	-50	0.11	9.5×10^6	IS
H5	255	350	49	1.92	0.31	295	- 70	0.06	4.3×10^8	CS
H6	304	400	35	1.45	0.25	382	- 60	0.06	3.5×10^8	CS
H7	351	454	32	1.38	0.22	395	- 40	0.61	2.3×10^4	CS
H8	258	360	49	2.35	0.28	345	- 70	0.06	4.6×10^8	IS
H9	311	408	44	1.98	0.21	389	- 60	0.05	3.3×10^8	IS
H10	330	445	38	1.82	0.2	422	- 50	0.09	9.5×10^7	IS
H11	264	364	47	1.91	0.22	350	- 70	0.06	4.7×10^8	CS
H12	245	350	50	1.85	0.28	338	- 80	0.06	4.5×10^8	IS
H13	253	355	49	1.83	0.29	342	- 80	0.07	2.5×10^8	IS
H14	293	405	45	1.65	0.21	390	- 60	0.06	4.0×10^8	IS
H15	355	453	38	1.51	0.22	435	- 60	0.09	9.1×10^6	IS
H16	234	342	52	1.85	0.33	275	- 80	0.09	4.2×10^6	CS
H17	308	412	36	1.48	0.21	398	- 70	0.09	3.2×10^6	CS
H18	335	448	34	1.38	0.57	380	- 60	0.51	9.3×10^4	CS
H19	240	352	50	2.28	2.9	335	- 80	0.05	8.2×10^8	IS
H20	303	410	44	1.88	2.1	387	- 60	0.06	4.5×10^8	IS
H21	359	460	37	1.7	2.0	437	- 60	0.08	4.1×10^6	IS
H22	252	359	50	1.86	2.2	339	- 80	0.07	4.5×10^6	CS
H23	250	355	48	1.89	0.28	335	- 80	0.06	9.5×10^8	IS
H24	245	355	47	1.85	0.27	348	- 80	0.06	6.5×10^8	IS
H25	288	395	46	1.69	0.25	375	- 60	0.07	2.1×10^8	IS
H26	348	443	37	1.54	0.21	420	- 60	0.09	7.5×10^6	IS
H27	244	345	46	1.88	0.36	283	- 80	0.09	5.2×10^8	CS
H28	297	402	33	1.45	0.21	365	- 70	0.09	3.2×10^8	CS
H29	345	454	33	1.36	0.47	385	- 60	0.51	9.3×10^4	CS
H30	252	358	48	2.15	0.24	330	- 80	0.07	8.3×10^8	IS
H31	292	390	43	1.92	0.2	372	- 60	0.09	3.2×10^8	IS
H32	343	448	38	1.72	0.18	421	- 60	0.07	7.5×10^6	IS
H33	251	357	47	1.79	0.2	341	- 80	0.06	6.5×10^6	CS

Note: YS = Yield strength, TS = Tensile strength, El = Elongation, r-value: Plasticity-anisotropy index, Δr -value: In-plane anisotropy index, PBYS = Post-bake yield strength, DBTT = ductility-brittleness transition temperature for investigating secondary work embrittlement, AS = Average size of precipitates, PN = The number of precipitates, IS = Steel of the invention, CS = Comparative steel

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[Example 3-3] MnCu-precipitated steel with AlN precipitation strengthening

In order to provide MnCu-precipitated steel with AlN precipitation strengthening according to the present invention, after steel slabs shown in Table 17 were reheated to a temperature of 1,200 °C, followed by finish rolling the steel slabs in order to provide hot rolled steel sheets, the hot rolled steel sheets were cooled at a cooling rate of 400 °C/min, and then coiled at 650 °C. Then, the hot rolled steel sheets were cold rolled at a reduction rate of 75 %, followed by continuous annealing the cold rolled steel sheets. The finish rolling was performed at 910 °C, which is above the A_{r3} transformation temperature, and the continuous annealing was performed by heating the steel sheets to 750 °C at a velocity of 10 °C/second for 40 seconds.

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Table 17

Sample No.	Component (wt%)										
	C	Mn	P	S	Al	N	Cu	Mo	R-4	R-5	R-2
	0.003-0.005	0.03-0.2	0.03-0.06	0.003-0.025	0.01-0.1	0.005-0.02	0.01-0.2	0.01-0.2	≤0.3	2-20	1-5
I1	0.0042	0.07	0.038	0.02	0.032	0.0085	0.03	-	0.1	2.5	1.96
I2	0.0038	0.1	0.042	0.015	0.042	0.0072	0.03	-	0.13	4.33	3.03
I3	0.0045	0.14	0.037	0.015	0.055	0.0092	0.05	-	0.19	6.33	3.11
I4	0.0045	0.2	0.05	0.009	0.07	0.008	0.05	-	0.25	13.9	4.55
I5	0.0015	0.17	0.04	0.012	0.042	0.0072	0.05	-	0.22	9.17	3.03
I6	0.0062	0.15	0.038	0.015	0.038	0.0014	0.12	-	0.27	9	14.1
I7	0.0036	0.25	0.042	0.009	0.04	0.0083	0.2	-	0.45	25	2.51
I8	0.0035	0.09	0.04	0.012	0.052	0.0093	0.043	0.019	0.13	5.54	2.91
I9	0.0046	0.11	0.039	0.011	0.053	0.011	0.053	0.082	0.16	7.41	2.51
I10	0.0038	0.12	0.042	0.012	0.061	0.012	0.085	0.16	0.21	8.54	2.64

I11	0.004	0.12	0.045	0.01	0.059	0.0095	0.065	0.26	0.19	9.25	3.23
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Note: R-2 = 0.52*Al/N, R-4 = Mn+Cu, R-5 = 0.5*(Mn+Cu)/S

Table 18

Sample No.	Mechanical Properties							AS (μm)	PN (number/ mm^2)	Remarks
	YS (Mpa)	TS (MPa)	El (%)	r-value (rm)	Δ r- value (Δ r)	PBYS (MPa)	DBTT ($^{\circ}\text{C}$)			
I1	250	355	48	1.89	0.28	343	- 80	0.06	9.5×10^8	IS
I2	245	355	47	1.85	0.27	348	- 80	0.06	6.5×10^8	IS
I3	248	352	47	1.89	0.23	345	- 80	0.07	2.1×10^8	IS
I4	254	348	45	1.84	0.28	330	- 60	0.09	7.5×10^6	IS
I5	240	342	46	1.88	0.32	280	- 80	0.09	5.2×10^6	CS
I6	247	362	40	1.55	0.38	335	- 70	0.09	3.2×10^6	CS
I7	253	352	42	1.66	0.37	295	- 60	0.51	9.3×10^4	CS
I8	247	350	49	2.11	0.25	339	- 80	0.06	8.9×10^8	IS
I9	251	359	48	2.13	0.24	340	- 80	0.06	7.5×10^6	IS
I10	245	348	49	2.15	0.22	332	- 80	0.08	1.1×10^8	IS
I11	250	352	47	1.80	0.21	335	- 60	0.09	1.5×10^6	CS

Note: YS = Yield strength, TS = Tensile strength, El = Elongation, r-value: Plasticity-anisotropy index, Δ r-value: In-plane anisotropy index, PBYS = Post-bake yield strength, DBTT = ductility-brittleness transition temperature for investigating secondary work embrittlement, AS = Average size of precipitates, PN = The number of precipitates, IS = Steel of the invention, CS = Comparative steel

Although the preferred embodiments of the present invention have been disclosed for illustrative purposes, those skilled in the art will appreciate that various modifications, additions and substitutions are possible, without departing from the scope and spirit of the invention as disclosed in the accompanying claims.